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QUARTERLY PROGRESS REPORT NO. 2

CRACK INITIATION IN METALLIC MATERIALS

FOR

Department of the Navy
Bureau of Naval Weapons
Washington 25, D. C.

May 1963



Metallurgical Research Laboratories
Department of Chemical Engineering and Metallurgy
Syracuse University

MET. E. 1040-630502

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Contract No. N600(19)-59514

by

K. Schroder, P. Packman and V. Weiss

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Department of Chemical Engineering and Metallurgy
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Introduction

The experimental investigation on the fracture properties of tungsten has been continued. The range of K_t values under study has been increased.

It was found that the surface preparation of recrystallized specimens influenced the extent of plastic deformation and fracture strength. Fracture strength σ_N increased with increasing plasticity. The lowest fracture strengths were observed for specimens with electrolytically produced notches, the highest for machined and electropolished specimens. The extrapolated ultimate fracture strength for all conditions studied is approximately constant, about 120 KSI.

Specimens tested in the as-received condition showed a similar low ultimate fracture strength only if tested in a liquid N_2 bath. Room temperature tests on specimens having both machined or electropolished notches, showed little plasticity and much higher extrapolated ultimate fracture strengths (190 to 220 KSI).

An attempt has been made to obtain a more detailed understanding of the dislocation mechanisms which are responsible for initiation and arrest of microcracks (see Appendix). It is suggested that two stages exist in the formation of microcracks. If a primary microcrack is formed, the stress concentration at its tip is unsymmetrical to the crack plane, if the crack plane is inclined to the load axis. This may finally lead to a secondary microcrack in a plane nearly perpendicular to the load axis.

Experimental Work

Tungsten Studies

Tests reported in the Final Report (1962) and the Quarterly Progress Report No. 1 are continuing. The data obtained are shown in Figs. 1 and 2.

Figure 1 shows the results on tungsten specimens in the as received condition. They have been tested in three major groups:

Series B. Electropolished surface. At least 0.003" of material is removed from the surface

Series R. Machined specimens

Series N. Specimens with an electropolished surface, tested at liquid Nitrogen temperatures.

The data have been extended to cover larger ranges of K_t values as previously reported.

Figure 2 gives the notch strength values of specimens recrystallized at 2000°C (Annealing time: 1 hr.). The extrapolated ultimate fracture strength for all specimens is about 120 KSI. Machined and electropolished specimens reached the highest notch strength. The absolute value slope of $\log \sigma_n$ versus $\log K_t$ for these specimens is smaller than those for specimens with notches produced with the electrolytic saw, indicating a higher amount of micro-plasticity at the notch root in the machined and electropolished specimens. Surprisingly enough, the notch strength at constant K_t of specimens with machined notches is in between these two curves.

Similar values to those obtained on recrystallized specimens of notch strength and extrapolated ultimate fracture strength were obtained in as-

received specimens, tested at liquid N₂ temperatures. Room temperature tests give a higher extrapolated ultimate fracture stress of about 220 KSI, with a microplasticity about as low as observed on recrystallized specimens with notches produced with an electrolytic saw.

As reported previously, attempts were made to vary the grain size of the tungsten specimens to test the validity of the $d^{1/2}$ grain size dependency of fracture strength as proposed by Cottrell. Several samples were recrystallized at 2000°C for one hour. The samples were then strained approximately 6% at 600°C in tension and then heat treated at 2000°C for 3 hours. They were then electropolished and etched in a 2% NaOH bath. Examination revealed that there was an increase in grain size. Typical photographs are shown in Figure 3. The ASM grain size has increased from about 5 to about 3.

Some progress can be reported in the study of the surface of notches by replica techniques. Several replicas of the surface of the notch have been produced that did not have the large amounts of distortion reported in the first progress report. There was no visible distortion of the replica when compared with the original, but there was some slight variation in the dimensions of the notch when examined at 47.5 magnifications. Differences of dimensions of specimen and replica are about 2 - 3% in notches about 0.012" deep and 0.006", see Figure 4.

A series of piggy-back notched specimen schematically shown in Figure 5 were made. These were electropolished specimens with notches of 0.012 and 0.023" radius provided with an additional notch of 0.0015" radius (depth: 0.001") put in at the base of the first notch with a diamond indenter.

These specimens will then be recrystallized and tested to see what effect the second notch has on the notch strength of the tungsten.

Remarks on the dislocation theory of crack initiation.

In order to develop a model for brittle fracture in LiF single crystals, it is important to consider three major aspects of crack initiation and cleavage. These are:

1. Dislocation mechanisms of slip that lead to the formation of an initial microcrack.
2. Growth of the initial microcrack (primary microcrack)
3. Nucleation of a secondary microcrack (see Appendix)
4. Growth of the primary and secondary microcrack at stresses lower than that required to produce brittle fracture.
(The length is below that of the Griffith crack).

It has been shown in the appendix that microcracks orientated at angles other than normal to the applied stress can produce finite cracks in the material that are visible as cracks in planes normal to the applied stresses. This implies that dislocation mechanisms which form a microcrack normal to the direction of the applied stress need not be the only mechanisms at work.

An analysis of several dislocation mechanisms for crack initiation is underway. Any dislocation model that considers the elastic stress produced by a pile-up of like dislocations is likely to overestimate the repulsive forces between dislocations because most models of close interaction of two dislocations do not consider non-linear elastic terms. The cal-

culated repulsive forces between them tend to become too large over very short distances of the order of a Burger's vector.

Stroh (1) considers for instance the piling-up of a series of dislocations of the same sign against a barrier. Utilizing the results of Eshelby, Frank and Nabarro (2) that the equilibrium distances between the dislocations are the zero's of the first derivative of N th Laguerre polynomial (n is the number of piled-up dislocations) he concluded that the stress would be great enough to open up a crack in the material if n dislocations were piled-up under a shear stress σ_s such that

$$n \geq \frac{12\gamma}{\sigma_s b} \quad (1)$$

Gilman has obtained a fracture shear stress of LiF of 9×10^4 gr/cm².

With $\gamma = 1310$ ergs/cm² and $b = 10^{-8}$ cm we see that the number of dislocations required in the pile-up is approximately $n \approx 1700$, which seems unduly high.

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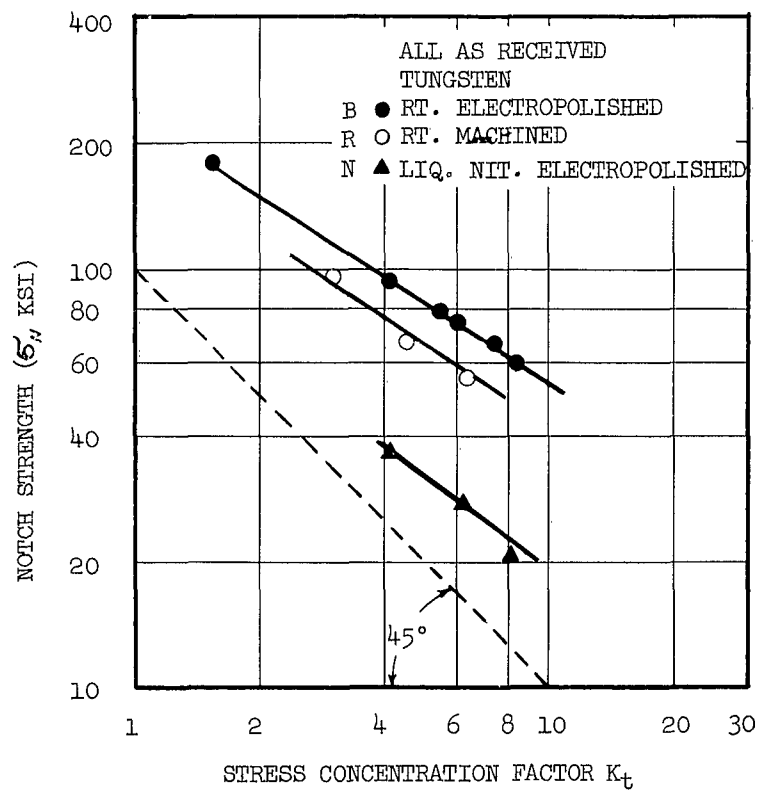


FIG. 1 NOTCHED TENSILE RESULTS ON TUNGSTEN SPECIMENS.

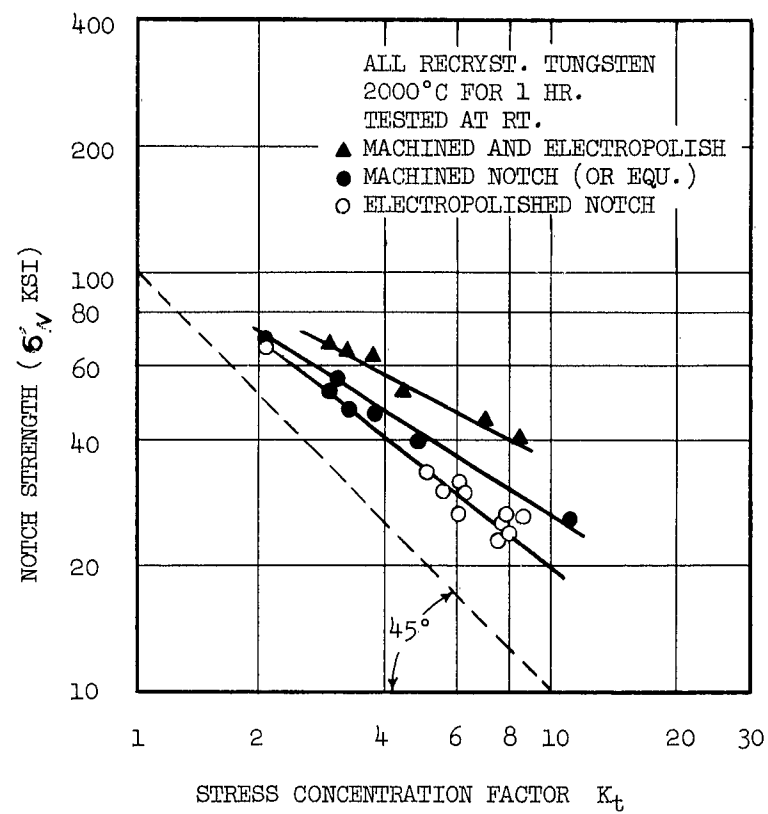


FIG. 2 NOTCHED TENSILE RESULTS ON RECRYSTALLIZED TUNGSTEN SPECIMENS.



FIG. 3a MICROPHOTOGRAPHS OF TUNGSTEN RECRYSTALLIZED AT 2000°C FOR 1 HOUR (100X)



FIG. 3b MICROPHOTOGRAPHS OF RECRYSTALLIZED TUNGSTEN STRAINED 6% AT 600°C AND HEAT TREATED AT 2000°C FOR 3 HOURS (100X)



FIG. 4a PHOTOGRAPH OF TUNGSTEN SPECIMEN WITH NOTCH AS PRODUCED BY ELECTROLYTIC SAW (47.5X)



FIG. 4b PHOTOGRAPH OF ANALDITE NO. 502 EPOXY REPLICA OF SAMPLE AT LEFT (47.5X)

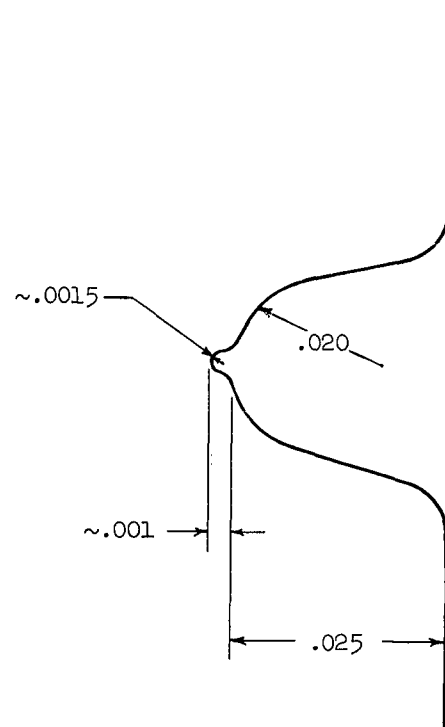


FIG. 5 SCHEMATIC SKETCH OF DOUBLE NOTCHED SPECIMENS.

SOME ASPECTS OF FORMATION, PROPAGATION,
AND ARREST OF MICROCRACKS IN METALS

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ABSTRACT

An attempt is made to show that all dislocation mechanism of crack formation proposed to date will lead to the formation of a microcrack which is approximately normal to the direction of the applied stress, even if the primary crack is inclined $40^\circ - 50^\circ$ to the stress field. This is due to the stress field at the tip of this primary initial crack which has a maximum stress concentration not at the center of the crack root but at the point where the tangent at the crack root is nearly parallel to the direction of the applied stress. This will lead to the formation of a secondary microcrack perpendicular to the direction of the applied stress as soon as a critical stress at the tip of the primary microcrack is reached.

It is further shown that several mechanisms exist which may increase the radius of curvature at the tip of the crack. It follows from Neuber's Theory of Stress Concentrations that this would be associated with a marked decrease of the stress concentration factor. These mechanisms may therefore be responsible for the arrest of a propagating crack and the experimentally observed larger stress requirement for crack propagation than for crack initiation.

1. INTRODUCTION

Several dislocation models have been proposed to explain crack initiation in crystals (Cottrell 1958, Fujita 1958, Gilman 1958, Orowan 1954, Stroh 1957 and 1958, Zener 1948). Except for the Cottrell mechanism, most models predict a crack either in or perpendicular to the stress axis. Because most slip planes are inclined approximately 40 to 50° to the stress axis, also most crack planes should be inclined roughly 40 to 50° to the stress axis.

However, experimental evidence shows that traces of cracks are most frequently perpendicular to the stress axis (Hahn et al. 1959). This does not necessarily prove that the crack plane itself is perpendicular to the stress axis. If the normal of the crack is inclined θ degrees to the stress axis (Fig. 1), the normal of this surface trace can include angles α with the stress axis which range from $0 \leq \alpha \leq \theta$ (Grewal 1963). A trivial calculation shows that one should expect the probability of having a certain $\tan \alpha$ value to be a sinusoidal function of β ($\tan \alpha = \tan \theta \cos \beta$, $0 \leq \beta \leq 180^\circ$). Using these relationships, one finds that the deviation from zero of the θ values is larger than the variation of α values as given by Hahn et al. (1959), but most θ values are close to 0° . On the other hand, Stroh (1959) showed that energy calculations make it rather unlikely that several pairs of $\langle 110 \rangle$ dislocations in different $\{111\}$ planes combine to dislocations with a $\langle 100 \rangle$ Burger's vector to form an obstacle as Cottrell assumes. Further, investigations on copper and other f.c.c. metals indicated that Cottrell's mechanism operates in these crystals where it is responsible for rapid work-hardening instead of forming cracks.

Most dislocation fracture mechanisms could operate in both b.c.c. and f.c.c. metals, but experimental evidence indicates that brittle fracture is found only in b.c.c. metals. Only a few experimental investigations on the dislocation structure of b.c.c. metals have been published. Brandon and Nutting (1959), using electron microscopy studies of iron, came to the conclusion that no lamella structure exists as in copper and that dislocations cross slip more easily in b.c.c. than in f.c.c. metals. Brittle fracture occurs in b.c.c. metals only below a critical temperature which is influenced by various parameters, but which is probably not directly related to a change in the glide system. There exists also conflicting evidence on the effect of dislocation densities on the ductility of tungsten. Experiments with high purity electropolished tungsten single crystals show plastic deformation at room temperature, but if deformed surface layers with high dislocation densities have not been removed, brittle fracture is observed. On the other hand, the ductile-brittle transition temperature in polycrystalline material is above 200°C (Atkinson 1960), yet, highly deformed polycrystalline tungsten with high dislocation densities show some ductile at room temperature. This seems to indicate that the dislocation density is not the critical parameter for the formation of cracks. It may have an influence on the yield stress of the system. This may be important in two respects; 1.) because a critical stress seems to be a criterion for brittle fracture, and 2.) a critical stress is required for the formation of twins which may be important for the nucleation of microcracks.

Microscopical investigations have shown that twinning is frequently associated with fracture initiation (Hull 1960, Sleeswyk 1962). Even if they are not visible in a light microscope, larger magnifications with an electron microscope may reveal submicroscopical twins. Twinning may be associated with the formation of microcracks due to inhomogeneous deformation. These twins may constitute obstacles for dislocation motion; such obstacles are needed in most dislocation mechanisms for crack initiation, except in the Cottrell and Fujita concept.

2. FORMATION OF PRIMARY AND SECONDARY MICROCRACKS

As mentioned above, energy considerations speak against the Cottrell mechanism. Also, that cross slip occurs more easily in b.c.c. metals than in f.c.c. metals, makes the Cottrell mechanism even less likely to initiate fracture in b.c.c. crystals compared with f.c.c. crystals.

Fujita's model (Fujita 1958) also seems to be applicable to both f.c.c. and b.c.c. crystals. Here, dislocation sources with different Burger's vectors operate in two lattice planes a few lattice constants apart, so that two opposite groups of dislocations form groups of one or several rows of vacancies (Fig. 2). Fujita found that a void created with such a model could be 10^{-5} to 10^{-6} cm wide and would constitute a primary crack with a stress concentration large enough to initiate fracture.

Fujita neglected the stress field of other dislocations. If two Frank-Read sources operate to originate dislocation groups for a Fujita crack, two pairs of dislocation groups are neglected which produce a compressive stress at the crack. This may be another reason why this model is inapplicable for the explanation of brittle fracture in metals.

Whatever model is finally accepted for the formation of primary microcracks, there is strong indication that such cracks will be 40 - 50° inclined to the direction of an applied tensile stress. Yet, Hahn et al. showed that all visible microcracks are approximately normal to the direction of the applied stress. Only Cottrell's and Fujita's models are at least partially in agreement with these observations.

Accepting a primary microcrack 40 - 50° inclined to the applied stress, it is possible to show that such cracks will cause the formation of secondary microcracks approximately normal to the direction of the applied stress. Fujita's approach can be easily applied to the microcrack models with 45° inclined crack planes. These microcracks may be approximated as very thin elliptical holes. Calculations by Neuber (1958) on the stress field of elliptical holes show that in the case of an approximately 45° inclined hole a maximum tensile stress is applied at the notch root at a point where the tangent is approximately parallel to the stress field (Fig. 3).

This can be shown as follows: the equation of the crack ellipse is given by Neuber (1957) as

$$\frac{x^2}{\sinh^2 u_0} + \frac{y^2}{\cosh^2 u_0} = 1 \quad (\text{Neuber, Eq. IV, 48}) \quad 1)$$

or

$$x = \sinh u_0 \cos v; \quad y = \cosh u_0 \sin v \quad (\text{Neuber, Eq. IV, 47}) \quad 2)$$

$$\sinh u_0 = (t/\xi - 1)^{-1/2} \quad (\text{Neuber, Eq. IV, 120}) \quad 3)$$

We assume that $\xi/t \ll 1$. ξ is the radius at the tip of the crack, $2t$ is the crack length ($\sqrt{\xi/t} = u_0 \ll 1$). The load per cross section perpendicular to the crack in the X direction (Fig. 3) is $p = P \cos \gamma$. According to Neuber, this gives a tangential stress $(\sigma'_v)_u = u_0$ at the surface of the elliptical crack equal to

$$(\sigma'_v)_u = u_0 = \frac{p}{2} \frac{\sinh 2u_0 - 1 - e^{2u_0} \cos 2v}{\sinh^2 u_0 + \cos^2 v} \quad (\text{Neuber, Eq. IV, 129}) \quad 4)$$

There is also a shear stress p'' at the crack in Y direction, which should produce a tangential stress $(\sigma''_v)_u = u_0$ at the elliptical surface of the crack given by:

$$(\sigma''_v)_u = u_0 = \frac{+3p''}{2} \frac{e^{2u_0} \sin 2v}{\sinh^2 u_0 + \cos^2 v} \quad (\text{Neuber, Eq. IV, 151}) \quad 5)$$

The net section stress terms p' and p'' are

$$P = \frac{p'}{\cos \gamma} = \frac{p''}{\sin \gamma} \quad 6)$$

γ = angle between crack plane and direction of applied load. Because the crack is assumed to be 45° inclined to the stress axis: $P/\sqrt{2} = p' = p''$. The stress σ at the crack surface is the sum of σ' and σ'' . Depending on the quadrant, their absolute values have to be added or subtracted (Fig. 3).

$$\sigma = \sigma' + \sigma'' \quad 7)$$

Neuber showed that the maximum of σ' is at the tip of the crack root, and the maximum of σ'' at

$$x = \xi \quad (\sqrt{\xi/t} = u_0 \ll 1) \quad (\text{Neuber, Eq. IV, 154}) \quad 8)$$

therefore, $\cos v$ is $\sqrt{t\xi} \ll t \sim 1$ and v is very close to $\pi/2$. It can be approximated by

$$v = \pi/2 - \alpha \quad 0 \leq \alpha \ll 1 \quad 9)$$

Therefore, we can set

$$\sinh 2u_0 \approx 2u_0; \quad e^{2u_0} \approx 1; \quad \cos 2v \approx -(1-2\alpha^2); \quad \sin 2v \approx 2\alpha; \quad \cos^2 v \approx \alpha^2 \quad 10)$$

Equations 4 to 10 give

$$\frac{\sigma}{P} \approx \sqrt{2} \frac{\sqrt{g/t} - 1/2 - 1/2(1-2\alpha^2) + 3/2\alpha}{g/t + \alpha^2}$$

At a maximum ($\alpha = \alpha_0$)

$$\frac{d\sigma}{d\alpha} = - \frac{d\sigma}{d\alpha} \text{ or } (g/t + \alpha_0^2) (-2\alpha_0 - 3) - (\sqrt{g/t} - \alpha_0^2 - 3\alpha_0) 2\alpha_0 = 0 \quad 11)$$

$2\alpha_0$ can be neglected compared with 3, and α_0^2 compared with $3\alpha_0$. The positive solution of this quadratic equation of α_0 is

$$\alpha_0 \approx \sqrt{g/t} \times 0.72 \quad 12)$$

From equations 1 and 2 follows that $\tan \beta$ at the ellipse (Fig. 3) is at the maximum of σ equal to

$$\tan \beta = \left| \frac{dx}{dy} \right| = \frac{x}{y(g/t)} = \frac{\sqrt{g/t} \alpha_0}{\sqrt{g/t} (1 - \frac{\alpha_0^2}{2})} \approx 0.72, \quad \beta \approx 36^\circ \quad 13)$$

It is the angle between the tangent at the maximum of σ at the ellipse and the Y-axis.

If a maximum stress criteria can be used for brittle crack initiation (Weiss 1960), a secondary microcrack will spread approximately perpendicular to the stress field from the primary microcrack. If one approximates a crack cross section by an ellipse 100 to 1000 Å long and about 10 Å thick, the stress concentration factor should be of the order of 20 to 100. This would be sufficient to start brittle fracture at stresses of $\frac{E}{200}$ to $\frac{E}{1000}$, with crack planes approximately perpendicular to the field of the outside applied load, as experimentally observed.

3. ARREST OF PROPAGATING CRACKS

It has been frequently found that it is easier to form microcracks than to propagate them. This seems to be difficult to reconcile with the

Griffith-Orowan fracture criterion $\sigma_G = \sqrt{1.4 E / 2t}$, (σ_G = critical fracture strength; $2t$ = crack length). These theories implicitly assume that the crack geometry does not change during crack growth; the Neuber theory shows that this would correspond to a constant radius curvature at the tip of the crack ($SCF \sim \sqrt{t/\rho}$). But this may not be justified; it is, for instance, incorrect if the crack reaches a grain boundary (Fig. 4). The effective radius at the grain boundary should be much larger than at the tip of the original crack. Further, it is not very likely that the orientation relation between these two grains makes it easy for the crack to pass the boundary.

Another arresting mechanism would be analog to the jog formation of a moving edge dislocation which cuts screw dislocations. If a crack propagates in a $\{100\}$ plane under an applied stress perpendicular to this plane and its front passes a screw dislocation with a Burger's vector perpendicular to the crack plane, it will split and move in two parallel $\{100\}$ planes, one Burger's vector apart. The connecting "crack jog" has an orientation which would not interact with the outside stress field. A single screw dislocation would have only a negligible effect on the movement of a crack. Subgrain boundaries which have a high screw dislocation density (twist boundaries) should have a much stronger effect and may account for both deceleration and branching of propagating cracks. To estimate this effect, one could assume that the increase in energy required to move the crack is proportional to the number of screw dislocations per atom number along the crack length. Thus for one screw dislocation per 5 atoms, which would correspond to a highly twisted subgrain boundary, the stress increase required to move the crack with constant speed would be about 20%.

Calculations show that a maximum shear stress due to a crack is expected in a plane at the crack tip, perpendicular to the crack plane (Williams 1961). A rather simple model shows (Fig. 5) that an edge dislocation source in this plane may easily absorb some of the energy of the system and, therefore, relieve some of the stress of the applied field.

Probably the most effective way to stop the propagation of a crack is achieved by increasing the radius of curvature at the tip of a crack. Neuber's theory shows that the stress concentration factor for thin elliptical holes is proportional to $\sqrt{t/\rho}$ where $2t$ is the crack length and ρ the effective crack radius at the tip. (We define as the "effective radius" a radius which would give the right value of the SCF $\sim \sqrt{t/\rho}$). For a crack which is propagating in a crystallographic plane, this radius ρ should be of the order $1/2$ atomic distance, this is $\approx 1 \text{ \AA}$. As mentioned above a maximum shear stress is produced at the tip of a crack, perpendicular to the crack plane. Edge dislocations with the correct sign in a suitable glide plane will move into the tip of the crack. Fig. 5 gives a schematic picture of possible changes of the crack tip geometry.

At the moment it is not possible to calculate the stress concentration factors of such crack tip geometries. But, it seems to be plausible to assume that the effective radius ρ increases markedly. This would be associated with a sharp drop in the stress at the tip of the crack and the crack stops growing.

Fig. 6 shows etchpit pattern at the tip of the crack in lithium fluoride crystals. It has been shown that these etchpits appear where

dislocations reach the surface. These micrographs prove that dislocations move into the tip of the crack as required in the theory above .

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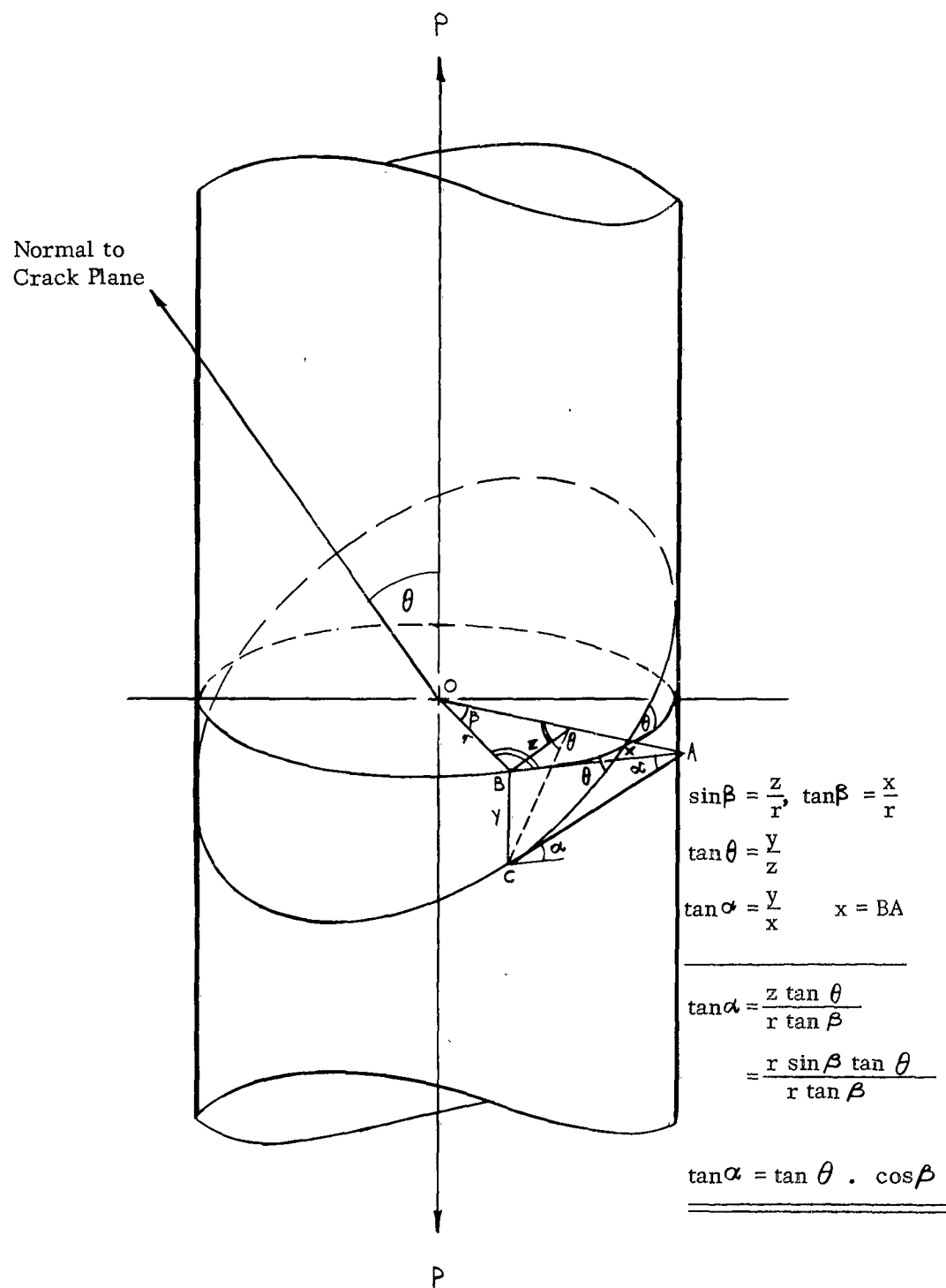


FIG. 1 ORIENTATION RELATION BETWEEN A CRACK AND ITS SURFACE TRACE IN A CYLINDRICAL ROD

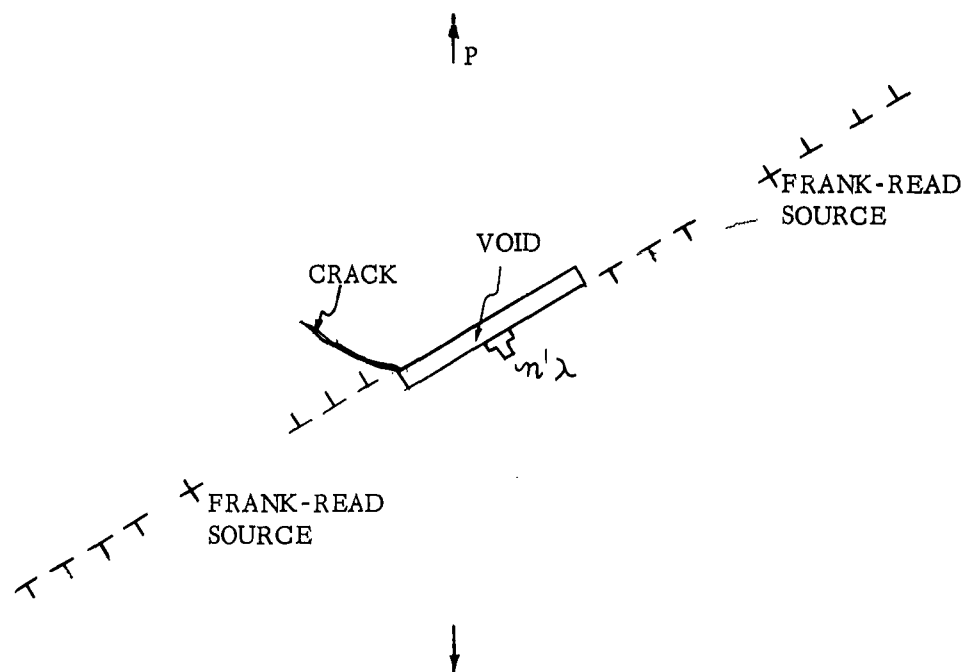


FIG. 2 FORMATION OF A PRIMARY MICROCRACK BY PARALLEL ROWS OF EDGE DISLOCATIONS, AND FORMATION OF A SECONDARY CRACK AT THE TIP OF THE PRIMARY CRACK (FUJITA, 1958)

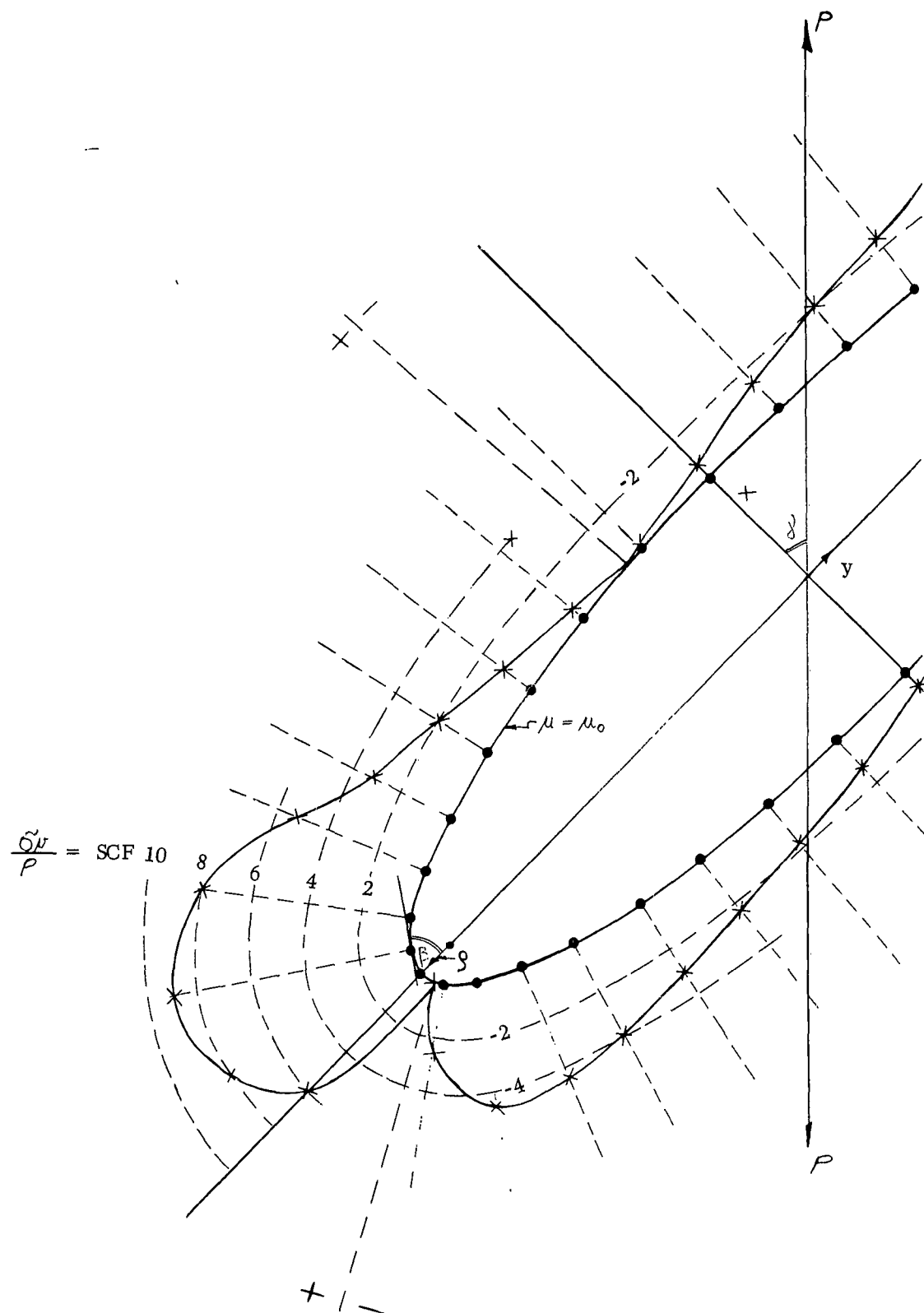


FIG. 3 TANGENTIAL STRESS AT THE SURFACE OF AN ELLIPTICAL CRACK WITH $t/\delta = 10$, 45° INCLINED TO THE AXIS OF AN APPLIED LOAD P

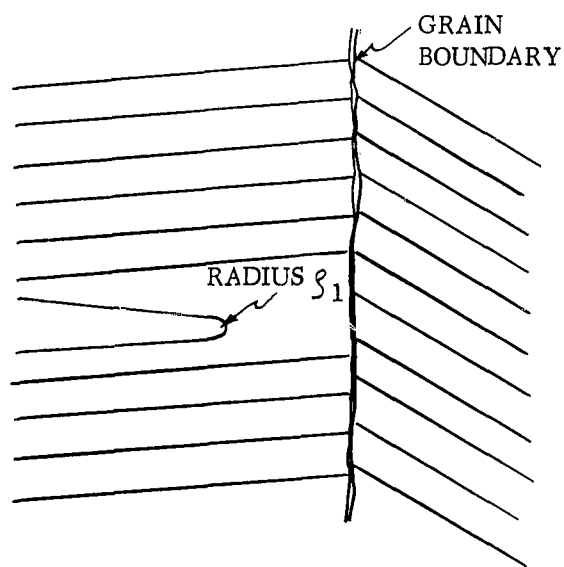


FIG. 4a CRACK BEFORE REACHING
A GRAIN BOUNDARY

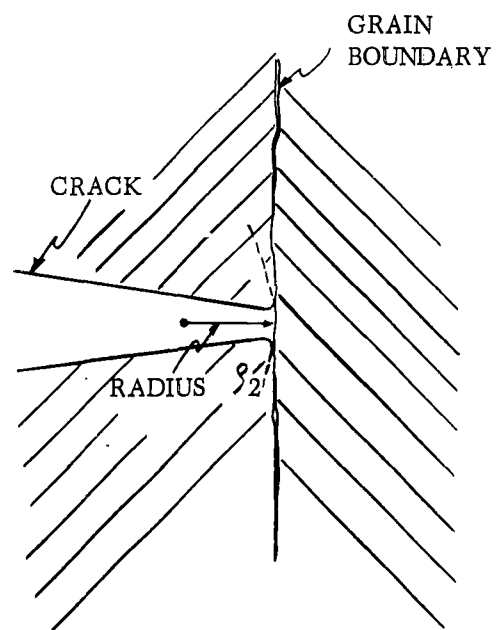


FIG. 4b CRACK AFTER REACHING
A GRAIN BOUNDARY

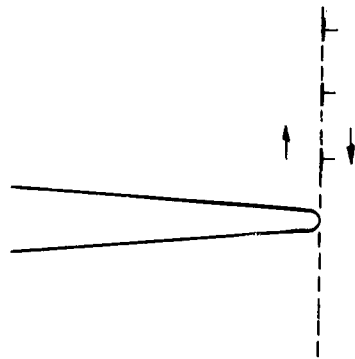


FIG. 5a CRACK BEFORE
ABSORBING EDGE
DISLOCATIONS AT
ITS TIP

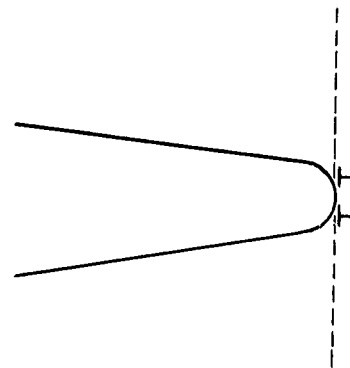


FIG. 5b CRACK AFTER
ABSORBING EDGE
DISLOCATIONS AT
ITS TIP

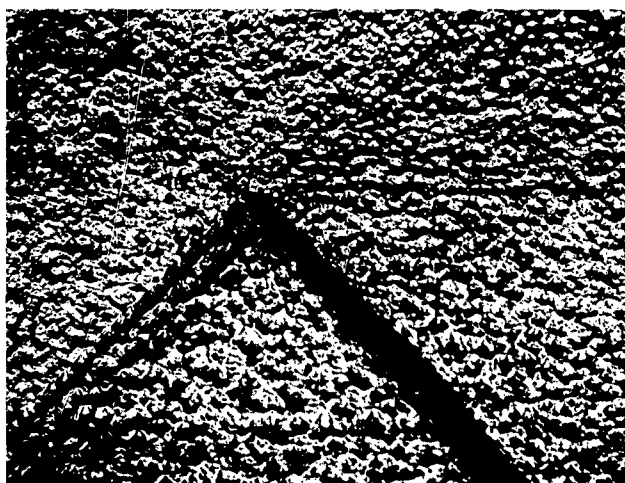


FIG. 6 ETCHPIT PATTERN NEAR CRACKS IN LITHIUM
FLUORIDE SINGLE CRYSTALS

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Boeing Airplane Company, Aerospace Division, P. O. Box 3825, Seattle 24,
Washington

General Dynamics Corporation, 3465 Wellington Road, Fort Worth, Texas
Attn: Mr. Charles W. Rogers

Armed Services Technical Information Agency, Arlington Hall Station,
Arlington, Virginia (Forwarded by separate letter. "Qualified Requestors
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Thiokol Chemical Corporation, Wasatch Division, Brigham City, Utah
Attn: Materials Development

Aluminum Company of America, Research Laboratories, New Kensington,
Pennsylvania

Defense Materials Information Center, 505 King Avenue, Columbus 1, Ohio

Thompson-Ramo Wooldridge, Inc., 23555 Euclid Avenue, Cleveland 17, Ohio
Attn: Dr. E. A. Steigerwald

Aerojet-General Corporation, Research and Engineering Department, Structural
Materials Division, Azusa, California

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